

Europäisches Patentamt
European Patent Office
Office européen des brevets



(11) **EP 1 291 448 B1**

(12) **EUROPEAN PATENT SPECIFICATION**

(45) Date of publication and mention
of the grant of the patent:
28.06.2006 Bulletin 2006/26

(51) Int Cl.:
C22C 38/00 (2006.01) C22C 38/06 (2006.01)
C22C 38/58 (2006.01) C21D 9/46 (2006.01)
C22C 38/04 (2006.01) C21D 8/02 (2006.01)

(21) Application number: **01906128.2**

(86) International application number:
PCT/JP2001/001004

(22) Date of filing: **14.02.2001**

(87) International publication number:
WO 2001/090431 (29.11.2001 Gazette 2001/48)

(54) **COLD ROLLED STEEL SHEET AND GALVANIZED STEEL SHEET HAVING STRAIN AGING
HARDENING PROPERTY AND METHOD FOR PRODUCING THE SAME**

**KALTGEWALZTES STAHLBLECH UND GALVANISIERTES STAHLBLECH MIT GUTEN
RECKALTERUNGSEIGENSCHAFTEN UND HERSTELLUNGSVERFAHREN DAFÜR**

**TOLE D'ACIER LAMINÉE A FROID ET TOLE D'ACIER GALVANISÉE POSSEDANT DES
PROPRIETES DE DURCISSEMENT PAR ECROUISSAGE ET PAR PRECIPITATION ET PROCEDE
DE PRODUCTION ASSOCIE**

(84) Designated Contracting States:
BE DE FR GB IT

(74) Representative: **Grünecker, Kinkeldey,
Stockmair & Schwanhäusser
Anwaltssozietät
Maximilianstrasse 58
80538 München (DE)**

(30) Priority: **26.05.2000 JP 2000156274**
28.06.2000 JP 2000193717
27.10.2000 JP 2000328924
02.11.2000 JP 2000335803

(56) References cited:
EP-A- 0 429 094 EP-A- 0 510 718
EP-A- 0 608 430 EP-A- 0 612 857
EP-A- 0 659 890 EP-A- 0 769 565
EP-A- 0 943 696 EP-A- 1 002 884
EP-A1- 0 608 430 JP-A- 6 116 682
JP-A- 7 034 136 JP-A- 8 035 039
JP-A- 11 279 691 JP-A- 58 042 753
JP-B2- 6 102 816 US-A- 4 838 955
US-A- 5 123 969 US-A- 5 405 463
US-A- 5 582 658 US-A- 5 690 755

(43) Date of publication of application:
12.03.2003 Bulletin 2003/11

(60) Divisional application:
04023082.3 / 1 498 506
04023101.1 / 1 498 507

(73) Proprietor: **JFE Steel Corporation**
Tokyo (JP)

(72) Inventors:
• **Kami, Chikara**
Technical Research Laboratories
Chiba-shi, Chiba 260-0835 (JP)
• **Tosaka, Akio**
Chiba Works
Chiba-shi, Chiba 260-0835 (JP)
• **Yamazaki, Takuya**
Technical Research Laboratories
Chiba-shi, Chiba 260-0835 (JP)

- **PATENT ABSTRACTS OF JAPAN vol.011, no. 136**
(C-419), 30 April 1987 (1987-04-30) -& JP 61
272323A (KAWASAKI STEEL CORP), 2 December
1986 (1986-12-02)
- **PATENT ABSTRACTS OF JAPAN vol.005, no.012**
(C-040), 24 January 1981 (1981-01-24) -& JP 55
141526 A (KAWASAKI STEEL CORP), 5
November 1980 (1980-11-05)
- **PATENT ABSTRACTS OF JAPAN vol.009, no. 184**
(C-294), 30 July 1985 (1985-07-30) -& JP 60 052528
A (KAWASAKI SEITETSU KK), 25 March 1985
(1985-03-25)

Note: Within nine months from the publication of the mention of the grant of the European patent, any person may give notice to the European Patent Office of opposition to the European patent granted. Notice of opposition shall be filed in a written reasoned statement. It shall not be deemed to have been filed until the opposition fee has been paid. (Art. 99(1) European Patent Convention).

EP 1 291 448 B1

- PATENT ABSTRACTS OF JAPAN vol. 016, no. 283 (C-0955), 24 June 1992 (1992-06-24) -& JP 04 074824 A (SUMITOMO METAL IND LTD), 10 March 1992 (1992-03-10)
- PATENT ABSTRACTS OF JAPAN vol. 009, no. 306 (C-317), 3 December 1985 (1985-12-03) -& JP 60 145355 A (KAWASAKI SEITETSU KK), 31 July 1985 (1985-07-31)
- PATENT ABSTRACTS OF JAPAN vol. 1995, no. 07, 31 August 1995 (1995-08-31) -& JP 07 090482 A (KAWASAKI STEEL CORP), 4 April 1995 (1995-04-04)

Description

Technical Field

[0001] The present invention relates to a cold-rolled steel sheet, which is suitable as raw material steel sheet for molded products such as building members, mechanical structural parts, automobile structural parts, etc., which is used at positions required to have structural strength, particularly, strength and/or stiffness in deformation, and which is subjected to heat treatment for increasing strength after processing such as pressing or the like.

1. In the present invention, "excellent strain age hardenability" means that in aging under conditions of holding at a temperature of 170°C for 20 min. after pre-deformation with a tensile strain of 5%, the increment in deformation stress (represented by the amount of $BH = \text{yield stress after aging} - \text{pre-deformation stress before aging}$) after aging is 80 MPa or more, and the increment in tensile strength (represented by $\Delta TS = \text{tensile strength after aging} - \text{tensile strength before pre-deformation}$) after strain aging (pre-deformation + aging) is 40 MPa or more.

Background Art

[0002] In producing a press-molded product of a thin steel sheet, a process of coating and baking at lower than 200°C is used as a method in which a material having low deformation stress before press forming to facilitate press forming, and then hardened after press forming to increase the strength of a part. As a steel sheet for such coating and baking, a BH steel sheet has been developed.

[0003] For example, Japanese Unexamined Patent Application Publication No. 55-141526 discloses a method in which Nb is added according to the contents of C, N and Al of steel, $Nb/(\text{dissolved C} + \text{dissolved N})$ by at% is limited in a specified range, and the cooling rate after annealing is controlled to adjust dissolved C and dissolved N in a steel sheet. Also, Japanese Examined Patent Application Publication No. 61-45689 discloses a method in which baking hardenability is improved by adding Ti and Nb.

[0004] However, in order to improve deep drawability, strength of the raw material sheets of the above-described steel sheets is decreased, and thus the steel sheets are not always sufficient as structural materials.

[0005] Japanese Unexamined Patent Application Publication No. 5-25549 discloses a method in which baking hardenability is improved by adding W, Cr and Mo to steel singly or in a combination.

[0006] In the above-described conventional techniques, strength is increased by bake-hardening due to the functions of small amounts of dissolved C and dissolved N in a steel sheet, and it is well known that a BH (Bake-Hardening) steel sheet is used for increasing only the yield strength of a material, not for increasing tensile strength. Therefore, the conventional techniques have only the effect of increasing the deformation start stress of a part, and the effect of increasing stress (tensile strength after forming) required for deformation over the entire deformation region from the deformation start to the deformation end is not said to be sufficient.

[0007] As a cold-rolled steel sheet having tensile strength increased after forming, for example, Japanese Unexamined Patent Application Publication No. 10-310847 discloses an alloying hot-dip galvanized steel sheet having tensile strength increased by 60 MPa or more by heat treatment in the temperature region of 200 to 450°C.

[0008] This steel sheet contains, by mass%, 0.01 to 0.08% of C, and 0.01 to 3.0% of Mn, and at least one of W, Cr, and Mo in a total of 0.05 to 3.0%, and further contains at least one of 0.005 to 0.1% of Ti, 0.005 to 0.1% of Nb and 0.005 to 0.1% of V according to demand, and the microstructure of the steel is composed of ferrite or mainly composed of ferrite.

[0009] However, this technique comprises forming a fine carbide in the steel sheet by heat treatment after forming to effectively propagate a dislocation of stress applied during pressing, thereby increasing the amount of strain. Therefore, heat treatment must be performed in the temperature range of 220 to 370°C. There is thus the problem of a necessary heat treatment temperature higher than general bake-hardening temperatures.

[0010] Furthermore, it is a very important problem that the body weight of an automobile is decreased in relation to the recent regulation of exhaust gases due to global environmental problems. In order to decrease the body weight of an automobile, it is effective to increase the strength of the used steel sheet, i.e., use a high-tensile-strength steel sheet, thinning the steel sheet used.

[0011] An automobile part using a high-tensile-strength thin steel sheet must exhibit a sufficient property according to its function. The property depends upon the part, and examples of the property include dent resistance, static strength against deformation such as bending, twisting, or the like, fatigue resistance, impact resistance, etc. Namely, the high-tensile-strength steel sheet used for an automobile part is required to be excellent in such a property after forming. The properties are related to the strength of a steel sheet after forming, and thus the lower limit of strength of the high-tensile-strength steel sheet used must be set for achieving thinning.

[0012] On the other hand, in the process for forming an automobile part, a steel sheet is press-molded. If the steel sheet has excessively high strength in press forming, the steel sheet causes the following problems: (1) deteriorating

shape fixability; (2) deteriorating ductility to cause cracking, necking, or the like during forming; and (3) deteriorating dent resistance (resistance to a dent produced by a local compressive load) when the sheet thickness is decreased. These problems thus inhibit the extension of application of the high-tensile-strength steel sheet to automobile bodies.

[0013] As a means for overcoming the problems, a steel sheet composed of ultra-low-carbon steel is known as a raw material, for example, for a cold-rolled steel sheet for an external sheet panel, in which the content of C finally remaining in a solid solution state is controlled to an appropriate range. This type of steel sheet is kept soft during press forming to ensure shape fixability and ductility, and its yield stress is increased by utilizing the strain aging phenomenon which occurs in the step of coating and baking at 170°C for about 20 minutes after press forming, to ensure dent resistance. This steel sheet is soft during press forming because C is dissolved in steel, while dissolved C is fixed to a dislocation introduced in press forming in the coating and baking step after press forming to increase the yield stress.

[0014] However, in this type of steel sheet, the increase in yield stress due to strain age hardening is kept down from the viewpoint of prevention of the occurrence of stretcher strain causing a surface defect. This causes the fault that the steel sheet actually less contributes to a reduction in weight of a part.

[0015] On the other hand, a steel sheet composed of dissolved N to further increase the amount of bake-hardening, and a steel sheet provided with a composite structure composed of ferrite and martensite to further improve baking hardenability have been proposed for applications in which the appearance is not so important.

[0016] For example, Japanese Unexamined Patent Application Publication No. 60-52528 discloses a method of producing a high-strength steel thin sheet having good ductility and spot weldability, in which steel containing 0.02 to 0.15% of C, 0.8 to 3.5% of Mn, 0.02 to 0.15% of P, 0.10% or less of Al, and 0.005 to 0.025% of N is hot-rolled by coiling at a temperature of 550°C or less, cold-rolled, and then annealed by controlled cooling and heat treatment. A steel sheet produced by the technique disclosed in Japanese Unexamined Patent Application Publication No. 60-52528 has a mixed structure comprising a low-temperature transformation product phase mainly composed of ferrite and martensite, and having excellent ductility, and high strength is achieved by utilizing strain aging due to positively added N during coating baking.

[0017] Although the technique disclosed in Japanese Unexamined Patent Application Publication No. 60-52528 greatly increases yield stress YS due to strain age hardening, the technique less increases tensile strength TS. Also, this technique causes large variations in the increment in yield stress YS to cause large variations in mechanical properties, and thus it cannot be expected that a steel sheet can be sufficiently thinned for contributing to a reduction in weight of an automobile part, which is currently demanded.

[0018] Japanese Examined Patent Application Publication No. 5-24979 discloses a high-tensile-strength cold-rolled steel thin sheet having baking hardenability which has a composition comprising 0.08 to 0.20% of C, 1.5 to 3.5% of Mn, and the balance composed of Fe and inevitable impurities, and a structure composed of homogeneous bainite containing 5% or less of ferrite, or bainite partially containing martensite. The cold-rolled steel sheet disclosed in Japanese Examined Patent Application Publication No. 5-24979 is produced by quenching in the temperature range of 200 to 400°C in the cooling process after continuous annealing, and then slowly cooling to obtain a structure mainly composed of bainite and having a large amount of bake-hardening which is not obtained by a conventional method.

[0019] However, in the steel sheet disclosed in Japanese Examined Patent Application Publication No. 5-24979, yield strength is increased after coating and baking to obtain a large amount of bake-hardening which is not obtained a conventional method, while tensile strength cannot be increased. Therefore, in application to a strength member, improvements in fatigue resistance and impact resistance after forming cannot be expected. Therefore, there is a problem in which the steel sheet cannot be used for applications greatly required to have fatigue resistance and impact resistance, etc.

[0020] Also, Japanese Examined Patent Application Publication No. 61-12008 discloses a method of producing a high-tensile-strength steel sheet having a high r value. This method is characterized by annealing ultra-low-C steel used as a raw material in a ferrite-austenite coexistence region after cold rolling. However, the resultant steel sheet has a high r value and a high degree of baking hardenability (BH property), but the obtained BH amount is about 60 MPa at most. Also, the yield point of the steel sheet is increased after strain aging, but TS is not increased, thereby causing the problem of limiting application to parts.

[0021] Furthermore, the above-described steel sheet exhibits excellent strength after coating and baking in a simple tensile test, but produces large variations in strength during plastic deformation under actual pressing conditions. Therefore, it cannot be said that the steel sheet is sufficiently applied to parts required to have reliability.

[0022] With respect of a hot-rolled steel sheet among coating baked steel sheets for press molded products, for example; Japanese Examined Patent Application Publication No. 8-23048 discloses a method of producing a hot-rolled steel sheet which is soft during processing, and has tensile strength increased by coating and baking after processing to be effective to improve fatigue resistance.

[0023] In this technique, steel contains 0.02 to 0.13 mass % of C, and 0.0080 to 0.0250 mass % of N, and the finisher deliver temperature and the coiling temperature are controlled to leave a large amount of dissolved N in the steel, thereby forming a composite structure as a metal structure mainly composed of ferrite and martensite. Therefore, an increase

of 100 MPa or more in tensile strength is achieved at the heat treatment temperature of 170°C after forming.

[0024] Japanese Unexamined Patent Application Publication No. 10-183301 discloses a hot-rolled steel sheet having excellent baking hardenability and natural aging resistance, in which the C and N contents are limited to 0.01 to 0.12 mass % and 0.0001 to 0.01 mass %, respectively, and the average crystal grain diameter is controlled to 8 μm or less to ensure a BH amount of as high as 80 MPa or more, and suppress the Al amount to 45 MPa or less.

[0025] However, this steel sheet is a hot-rolled sheet, and is thus difficult to obtain a high r value because the ferrite aggregation texture is made random due to austenite-ferrite transformation. Therefore, the steel sheet cannot be said to have sufficient deep drawability.

[0026] Furthermore, even if the hot-rolled steel sheet obtained by this technique is used as a starting material for cold rolling and recrystallization annealing, the increase in tensile strength obtained after forming and heat treatment is not always equivalent to a hot-rolled steel sheet, and a BH amount of as high as 80 MPa or more cannot be always obtained. This is because the microstructure of the cold-rolled steel becomes different from that of hot-rolled one due to cold rolling and recrystallization annealing, and strain greatly accumulates during cold rolling to easily form a carbide, a nitride or a carbonitride, thereby changing the states of dissolved C and dissolved N.

[0027] Prior art EP 0 943 696 A1 relates to steel plates for drum cans and a method for manufacturing the same. This prior art method suggests to limit the N content to at most 0.005% by weight if the composition further contains Nb. Further, the Al content according to this prior art is at least 0.035%.

[0028] According to prior art JP-A- 61-27 23 23, a method for manufacturing blackplate for service treatment by continuous annealing is provided. The mandatory elements according to this prior art are C, Si, Mn, P, N and Al.

[0029] None of the above prior art documents, however, provides a cold-rolled steel sheet with satisfactory strain age hardenability properties.

[0030] In consideration of the above-described present conditions, an object of the present invention is to provide a cold-rolled steel sheet for deep drawing, which has excellent strain aging hardenability ($\text{BH} \geq 80 \text{ MPa}$ and $\Delta\text{T}_S \geq 40 \text{ MPa}$).

Disclosure of Invention

[0031] In order to achieve the object, the inventors produced various steel sheets having different compositions under various production conditions, and experimentally evaluated various material properties. As a result, it was found that both moldability and hardenability after forming can be improved by using as a strengthening element N, which has not been positively used before in a field requiring high processability, and effectively using the great strain age hardening phenomenon manifested by the action of the strengthening element.

[0032] The inventors also found that in order to advantageously use the strain age hardening phenomenon due to N, the strain age hardening phenomenon due to N must be advantageously combined with a condition for coating and baking an automobile, or further positively combined with a heat treatment condition after forming. It was thus found to be effective to appropriately control the hot rolling condition, the cold rolling and the cold rolling annealing condition to control the microstructure of a steel sheet and the amount of dissolved N in certain ranges. It was also found that in order to stably manifest the strain age hardening phenomenon due to N, it is important to control the Al content of the composition according to the N content.

[0033] A steel sheet of the present invention exhibits higher strength after coating and baking in a simple tensile test, as compared with a conventional steel sheet, and exhibits small variations in strength in plastic deformation under actual pressing conditions and stable part strength, thereby enabling application to parts required to have reliability. For example, a portion where large strain is applied to decrease the thickness has higher hardenability than other portions, and is considered homogeneous when being evaluated based on a surcharge load ability of (thickness) \times (strength), thereby stabilizing strength as a part.

[0034] As a result of further intensive research for achieving the objects, the inventors found the following:

1) In order to increase tensile strength after forming and heat treatment, a new dislocation must be introduced for progressing tensile deformation. The movement of the dislocation introduced by pre-deformation must be prevented by interaction between the dislocation introduced by forming and an interstitial element or a precipitate even when upper yield stress is attained.

2) In order to obtain the above interaction by forming a carbide, a nitride or a carbonitride of W, Cr, Mo, Ti, Nb, Al or the like, the heat treatment temperature after forming must be increased to 200°C or more. Therefore, it is more advantageous to positively use the interstitial element or a Fe carbide or Fe nitride because the heat treatment temperature after forming is decreased.

3) Of interstitial elements, dissolved N has the higher interaction with a dislocation introduced by forming than dissolved C even when the heat treatment temperature after forming is decreased, and thus a dislocation introduced by pre-deformation less moves when upper yield stress is attained.

4) Although dissolved N is present in crystal grains and crystal grain boundaries in steel, the increase in strength

after forming and heat treatment increases as the area of the crystal grain boundaries increases. Namely, the smaller crystal grain diameter is advantageous.

5) In order to increase the crystal grain boundary area, it is advantageous to add a combination of Nb and B and cool immediately after the end of hot rolling, suppressing normal grain growth of ferrite grains after the end of hot rolling and suppressing grain growth by recrystallization annealing after cold rolling.

[0035] The present invention has been achieved based on the above findings. The findings were obtained from the experiment described below.

Experiment 1

[0036] A sheet bar (thickness: 30 mm) having a composition containing, by mass %, 0.0015% of C, 0.0010% of B, 0.015 of Si, 0.5% of Mn, 0.03% of P, 0.08% of S and 0.011% of N, 0.005 to 0.05% of Nb and 0.005 to 0.03% of Al, and the balance composed of Fe and inevitable impurities was uniformly heated at 1150°C, hot-rolled by three passes so that the temperature of the final pass was 900°C higher than the A_{r3} transformation point, and then cooled with water for 0.1 second. Then, the sheet bar was subjected to heat treatment corresponding to coiling at 500°C for 1 hour.

[0037] The thus-obtained hot-rolled sheet having a thickness of 4 mm was cold-rolled with a rolling reduction ratio of 82.5%, recrystallized and annealed at 800°C for 40 seconds, and then temper-rolled with a rolling reduction ratio of 0.8%. Then, a tensile test specimen of JIS No. 5 was obtained from the resultant cold-rolled sheet in the rolling direction, and tensile strength was measured with a strain rate of 0.02/s by using a general tensile testing machine. Also, tensile strain of 10% was applied to a tensile test specimen of JIS No. 5 separately obtained from the cold-rolled sheet in the rolling direction, and then the specimen was subjected to a normal tensile test after heat treatment at 120°C for 20 minutes. The difference between the tensile strength of the specimen obtained from the cold-rolled sheet and the tensile strength of the specimen heat treated at 120°C for 20 minutes after application of 10% tensile strain was considered as the increase in tensile strength after forming (ΔTS).

[0038] Fig. 1 shows the results of measurement of the relation between the steel compositions ($N\% - 14/93 \cdot Nb\% - 14/27 \cdot Al\% - 14/11 \cdot B\%$) and ΔTS .

[0039] The figure indicates that ΔTS becomes 60 MPa or more when the value of ($N\% - 14/93 \cdot Nb\% - 14/27 \cdot Al\% - 14/11 \cdot B\%$) satisfies 0.0015 mass %.

Experiment 2

[0040] A sheet bar (thickness: 30 mm) having a composition containing, by mass %, 0.0010% of C, 0.02 of Si, 0.6% of Mn, 0.01% of P, 0.009% of S and 0.012% of N, 0.01% of Al, 0.015% of Nb, 0.00005 to 0.0025% of B, and the balance composed of Fe and inevitable impurities was uniformly heated at 1100°C, hot-rolled by three passes so that the temperature of the final pass was 920°C higher than the A_{r3} transformation point, and then cooled with water for 0.1 second. Then, the sheet bar was subjected to heat treatment corresponding to coiling at 450°C for 1 hour.

[0041] The thus-obtained hot-rolled sheet having a thickness of 4 mm was cold-rolled with a rolling reduction ratio of 82.5%, recrystallized and annealed at 820°C for 40 seconds, and then temper-rolled with a rolling reduction ratio of 0.8%. Then, a tensile test specimen of JIS No. 5 was obtained from the resultant cold-rolled sheet in the rolling direction, and tensile strength was measured with a strain rate of 0.02/s by using a general tensile testing machine. Also, tensile strain of 10% was applied to a tensile test specimen of JIS No. 5 separately obtained from the cold-rolled sheet in the rolling direction, and then the specimen was subjected to a normal tensile test after heat treatment at 120°C for 20 minutes.

[0042] Fig. 2 shows the results of measurement of the relation between the B content of steel and ΔTS . This figure indicates that with a B content of 0.0005 to 0.0015 mass %, a high ΔTS of 60 MPa or more can be obtained.

[0043] As a result of observation of the microstructure, it was also found that by adding a combination of Nb and B to make fine crystal grains, a high ΔTS can be obtained.

[0044] Namely, with a B content of less than 0.0005 mass %, the effect of making fine crystal grains by adding a combination with Nb is small. On the other hand, with a B content of over 0.0015 mass %, the amount of B segregated in the grain boundaries and the vicinities thereof is increased to decrease the amount of effective dissolved N because of the strong interaction between B atoms and N atoms, thereby possibly decreasing ΔTS .

Experiment 3

[0045] A sheet bar (thickness: 30 mm) of each of steel A having a composition containing, by mass %, 0.0010% of C, 0.012% of N, 0.0010% of B, 0.01% of Si, 0.5% of Mn, 0.03% of P, 0.008% of S, 0.014% of Nb, 0.01% of Al, and the balance composed of Fe and inevitable impurities, and steel B having a composition containing, by mass %, 0.010% of C, 0.0012% of N, 0.0010% of B, 0.01% of Si, 0.5% of Mn, 0.03% of P, 0.008% of S, 0.014% of Nb, 0.01% of Al, and the

balance composed of Fe and inevitable impurities was uniformly heated at 1150°C, hot-rolled by three passes so that the temperature of the final pass was 910°C higher than the Ar_3 transformation point, and then cooled with a gas for 0.1 second. Then, each of the sheet bars was subjected to heat treatment corresponding to coiling at 600°C for 1 hour.

[0046] Each of the thus-obtained hot-rolled sheets having a thickness of 4 mm was cold-rolled with a rolling reduction ratio of 82,5%, recrystallized and annealed at 880°C for 40 seconds, and then temper-rolled with a rolling reduction ratio of 0.8%.

[0047] Then, a tensile test specimen of JIS No. 5 was obtained from each of the resultant cold-rolled sheets in the rolling direction, and tensile strength was measured with a strain rate of 0.02/s by using a general tensile testing machine. Also, tensile strain of 10% was applied to a tensile test specimen of JIS No. 5 separately obtained from each of the cold-rolled sheets in the rolling direction, and then the specimen was subjected to a normal tensile test after heat treatment at various temperatures for 20 minutes.

[0048] Fig. 3 shows the results of measurement of the influence of the heat treatment temperature after forming on ΔTS . This figure indicates that in the relatively low temperature region of heat treatment temperatures of 200°C or less after forming, the ultra-low carbon steel A having a high N content exhibits higher ΔTS than the semi-ultra low carbon steel B having a low N content, and while in the high temperature region, both steel materials exhibit substantially the same ΔTS . There experimental results reveal that in order to ensure ΔTS in the low temperature region, it is effective to use dissolved N.

[0049] Fig. 4 shows the results of measurement of the influences of the crystal grain diameter d and steel compositions (N% - 14/93•Nb% - 14/27•Al% - 14/11•B%) on a decrease (ΔEI) in elongation by natural aging and an increase in tensile strength (ΔTS) after forming. The decrease (ΔEI) in elongation was evaluated by the difference between the total elongation measured with the test specimen of JIS NO. 5 obtained from each of the cold-rolled sheets in the rolling direction, and the total elongation measured with the separately obtained test specimen after holding at 100°C for 8 hours for accelerating natural aging.

[0050] Fig. 4 indicates that when the value of (N% - 14/93•Nb% - 14/27•Al% - 14/11•B%) is 0.0015 mass % or more, and the crystal grain diameter d is 20 μm or less, both high ΔTS and low ΔEI can be achieved.

Experiment 4

[0051] A sheet bar of steel containing 0.0015% of C, 0.30 of Si, 0.8% of Mn, 0.03% of P, 0.005% of S and 0.012% of N, and 0.02 to 0.08% of Al was uniformly heated at 1050°C, hot-rolled by seven passes so that the temperature of the final pass was 670°C, and then recrystallized and annealed at 700°C for 5 hours. The thus-obtained hot-rolled sheet having a thickness of 4 mm was cold-rolled with a rolling reduction ratio of 82,5%, recrystallized and annealed at 875°C for 40 seconds, and then temper-rolled with a rolling reduction of 0.8%. Then, a tensile test specimen of JIS No. 5 was obtained from the resultant cold-rolled sheet in the rolling direction, and TS x r value and ΔTS were measured with a strain rate of 3×10^{-3} /s by using a general tensile testing machine. The results are shown in Fig. 5. In this figure, when $N/Al \geq 0.03$ is satisfied, TS x r value ≥ 750 and $\Delta TS \geq 40$ MPa are achieved. It was also confirmed that when $N/Al \geq 0.03$, BH ≥ 80 MPa is attained.

Experiment 5

[0052] A sheet bar of steel containing 0.0015% of C, 0.0010% of B, 0.01 of Si, 0.5% of Mn, 0.03% of P, 0.008% of S and 0.011% of N, 0.005 to 0.05% of Nb, and 0.005 to 0.03% of Al was uniformly heated at 1000°C, hot-rolled by seven passes so that the temperature of the final pass was 650°C, and then recrystallized and annealed at 800°C for 60 seconds. The thus-obtained hot-rolled sheet having a thickness of 4 mm was cold-rolled with a rolling reduction ratio of 82,5%, recrystallized and annealing at 880°C for 40 seconds, and then temper-rolled with a rolling reduction ratio of 0.8%. Then, a tensile test specimen of JIS No. 5 was obtained from the resultant cold-rolled sheet in the rolling direction, and TS x r value, BH and ΔTS were measured with a strain rate of 3×10^{-3} /s by using a general tensile testing machine. The relations between the measured values and $N/(Al+Nb+B)$ are shown in Fig. 5. In this experiment, steel containing 0.005 to 0.05% of Nb and 0.0010% of B was used. This figure indicates that in the range of $N/(Al+Nb+B) \geq 0.30$, BH ≥ 80 MPa, $\Delta TS \geq 60$ MPa, and TS x r value ≥ 850 are achieved.

Experiment 6

[0053] A sheet bar of steel containing 0.0010% of C, 0.02 of Si, 0.6% of Mn, 0.01% of P, 0.009% of S and 0.015% of N, 0.01% of Al, 0.015% of Nb and 0.0001 to 0.0025% of B was uniformly heated at 1050°C, hot-rolled by seven passes so that the temperature of the final pass was 680°C, and then recrystallized and annealed at 850°C for 5 hours. The thus-obtained hot-rolled sheet having a thickness of 4 mm was cold-rolled with a rolling reduction ratio of 82,5%, recrystallized and annealed at 880°C for 40 seconds, and then temper-rolled with a rolling reduction ratio of 0.8%. Then, a

tensile test specimen of JIS No. 5 was obtained from the resultant cold-rolled sheet in the rolling direction, and TS x r value, BH and ΔTS were measured with a strain rate of 3×10^{-3} /s by using a general tensile testing machine. The relations between the measured values and the B content are shown in Fig. 6.

[0054] This figure indicates that in the B content range of 0.0003 to 0.0015%, BH \geq 80 MPa, $\Delta TS \geq$ 60 MPa, which is higher than the case of B < 0.0003%, and TS x r value \geq 850 are achieved. As a result of observation of the microstructure, it was also confirmed that in this B range, crystal grains are significantly made fine.

[0055] The results shown in Figs. 5 and 6 indicate that in the range of $N/(Al+Nb+B) \geq 0.30$ wherein B \geq 0.0003%, the crystal grains are further made fine by combining Nb, and ΔTS and the level of TS x r value are further improved. When B < 0.0003%, the effect of making fine crystal grains by combining Nb is not exhibited. On the other hand, when B > 0.0015%, properties further deteriorate. This is possibly due to the fact that the amount of B segregated in the grain boundaries and the vicinities thereof is increased to decrease the amount of effective dissolved N due to the strong interaction between B and N atoms. The same research as described above was carried out for the case in which Ti and V were added in place of Nb, and it was confirmed that the same effect as Nb could be obtained. The present invention has been achieved based on the above-described findings, and the gist of the invention was follows.

[0056] According to the present invention, a cold-rolled steel sheet having excellent strain age hardenability comprises a composition, by mass %, comprising:

C: 0.15% or less;
Si: 0.005% to 1.0%;
Mn: 0.01% to 2.0%;
P: 0.1% or less;
Nb: 0.005% to 0.050%
B: 0.0001% to 0.0030%
S: 0.01% or less;
Al: 0.005 to 0.030%; and
N: 0.0050 to 0.0400%;

wherein N/Al is 0.30 or more, the amount of dissolved N is 0.0010% or more, and the balance is composed of Fe and inevitable impurities, and said steel sheet has a crystal grain diameter of 20 μm or less.

[0057] Further according to the invention the composition may optionally further comprise at least one of the following groups a to c:

Group a: at least one of Cu, Ni, Cr and Mo in a total of 1.0% or less;
Group b: one or both of Ti and V in a total of 0.1% or less; and
Group c: one or both of Ca and REM in a total of 0.0010 to 0.010%.

Brief Description of the Drawings

[0058]

Fig. 1 shows the relation between steel compositions ($N\% - 14/93 \cdot Nb\% - 14/27 \cdot Al\% - 14/11 \cdot B\%$) and the increase in tensile strength (ΔTS) after forming.

Fig. 2 shows the relation between the B content and ΔTS of steel containing a combination Nb and B.

Fig. 3 shows comparison of the difference in increase in tensile strength by heat treatment after forming in a low temperature region between steel B (conventional steel) containing a large amount of dissolved C and steel A (steel of this invention) containing a large amount of dissolved N.

Fig. 4 shows the influence of the crystal grain diameter d and steel compositions ($N\% - 14/93 \cdot Nb\% - 14/27 \cdot Al\% - 14/11 \cdot B\%$) on the decrease in elongation (ΔEl) due to natural aging and the increase in tensile strength (ΔTS) after forming.

Fig. 5 shows the relations between TS x r value, BH, ΔTS and $N/(Al+Nb+B)$.

Fig. 6 shows the relations between TS x r value, BH, ΔTS and the B amount.

Best Mode for Carrying Out the Invention

[0059] Description will now be made of the reasons for limiting compositions to the ranges below in accordance with a first embodiment of the present invention.

C: less than 0.15 mass %

[0060] From the viewpoint of excellent deep drawability and press moldability, C is advantageously as small as possible. Also, redissolution of NbC proceeds in the annealing step after cold rolling to increase the amount of dissolved C in crystal grains, thereby easily causing deterioration in natural aging resistance. Therefore, the C amount is preferably suppressed to less than 0.01 mass %, more preferably 0.0050 mass % or less, and most preferably 0.0030 mass % or less.

Si: 0.005 to 1.0 mass %

[0061] Si is a useful composition for suppressing a decrease in elongation, and improving strength. However, with a Si content of less than 0.005 mass %, the effect of addition of Si is insufficient, while with a Si content of over 1.0 mass %, surface properties deteriorate to deteriorate ductility. Therefore, the Si content is limited to the range of 0.005 to 1.0 mass %, and preferably the range of 0.01 to 0.75 mass %.

Mn: 0.01 to 2.0 mass %

[0062] Mn not only is useful as a strengthening composition for steel, but also has the function to suppress embrittlement with S due to the formation of MnS. However, with a Mn content of less than 0.01 mass %, the effect of addition of Mn is insufficient, while with a Mn content of over 2.0 mass %, surface properties deteriorate to deteriorate ductility. Therefore, the Mn content is limited to the range of 0.01 to 2.0 mass %, and preferably the range of 1.10 to 0.75 mass %.

P: 0.1 mass % or less

[0063] P is a solid solution strengthening element which effectively contributes to reinforcement of steel. However, with a P content of over 0.1 mass %, deep drawability deteriorates due to the formation of phosphide such as $(\text{FeNb})_x\text{P}$ or the like. Therefore, P is limited to 0.10 mass % or less.

S: 0.01 mass % or less

[0064] With a high S content, the amount of inclusions is increased to deteriorate ductility. Therefore, contamination with S is preferably prevented as much as possible, but an S content up to 0.01 mass % is allowable.

Al: 0.005 to 0.030 mass %

[0065] Al is added as a deoxidizer for improving the yield of carbonitride forming components. However, with an Al content of less than 0.005 mass %, the effect is insufficient, while with an Al content of over 0.030 mass %, the amount of N to be added to steel is increased to easily cause slab defects during steel making. Therefore, Al is contained in the range of 0.005 to 0.030 mass %.

N: 0.005 to 0.040 mass %

[0066] In the present invention, N is an important element which plays the role of imparting strain age hardenability to a steel sheet. However, with an N content of less than 0.005 mass %, a sufficient strain age hardenability cannot be obtained, while with an N content of as high as over 0.040 mass %, press moldability deteriorates. Therefore, N is contained in the range of 0.005 to 0.040 mass %, and preferably in the range of 0.008 to 0.015 mass %.

B: 0.0001 to 0.003 mass %

[0067] B is added in a combination with Nb to exhibit the function to effectively make fine the hot-rolled structure and the cold-rolled recrystallized annealed structure and to improve cold-work embrittlement resistance. However, with a B content of less than 0.0001 mass %, the sufficient effect of making fine the structures cannot be obtained, while with a B content of over 0.003 mass %, the amount of BN precipitate is increased, and dissolution in the slab heating step is hindered. Therefore, B is contained in the range of 0.0001 to 0.003 mass %, preferably in the range of 0.0001 to 0.0015 mass %, and more preferably in the range of 0.0007 to 0.0012 mass %.

Nb: 0.005 to 0.050 mass %

[0068] Nb is added in a combination with B to contribute to refinement of the hot-rolled structure and the cold-rolled

recrystallized annealed structure, and have the function to fix dissolved C as NbC. Furthermore, Nb forms a nitride NbN to contribute to refinement of the cold-rolled recrystallized annealed structure. However, with a Nb content of less than 0.005 mass %, not only it becomes difficult to precipitate and fix dissolved C, but also the hot-rolled structure and the cold-rolled, recrystallized, annealed structure are not sufficiently made fine, while with a Nb content of over 0.050 mass %, ductility deteriorates. Therefore, Nb is contained in the range of 0.005 to 0.050 mass %, and preferably 0.010 to 0.030 mass %.

[0069] As described above, Nb has the function to fix dissolved C as NbC, and forms a nitride NbN. Similarly, Al and B form AlN and BN, respectively. Therefore, in order to ensure the sufficient amount of dissolved N and sufficiently decrease the amount of dissolved C, it is important to satisfy the following relations (1) and (2):

$$N\% \geq 0.0015 + 14/93 \cdot Nb\% + 14/27 \cdot Al\% + 14/11 \cdot B\% \dots (1)$$

$$C\% \leq 0.5 \cdot (12/93) \cdot Nb\% \dots (2)$$

[0070] In the present invention, in order to obtain a high strain aging property and prevent aging deterioration, the crystal grain diameter is decreased to 20 μm or less.

[0071] Namely, as described above with reference to Fig. 4, even when $(N\% - 14/93 \cdot Nb\% - 14/27 \cdot Al\% - 14/11 \cdot B\%) \geq 0.0015$ mass %, i.e., when a relatively large amount of dissolved N is contained, ΔE_I can be suppressed to 2.0% or less by decreasing the crystal grain diameter d to 20 μm or less. The crystal grain diameter d is more preferably decreased to 15 μm or less. This is because, as shown in Fig. 4, ΔE_I can be suppressed to 2.0% or less by decreasing the crystal grain diameter d to 15 μm or less.

[0072] In the following an example of a method for manufacturing a cold-rolled steel sheet according to the above embodiment of the invention will be described.

[0073] Steel having the above-described suitable composition is melted by a known melting method such as a converter or the like, and a steel slab is formed by an ingot making method or a continuous casting method.

[0074] Then, the steel slab is heated and soaked, and then hot-rolled to form a hot-rolled sheet. In this example, the heating temperature of hot rolling is not specified, but the heating temperature of hot rolling is preferably set to 1300°C or less. This is because it is advantageous to fix and precipitate dissolved C as a carbide in order to improve deep drawability. In order to further improve processability, the heating temperature is preferably set to 1150°C or less. However, with a heating temperature of less than 900°C, improvement in processability is saturated to conversely increase the rolling load in hot rolling, thereby increasing the danger of causing a rolling trouble. Therefore, the lower limit of the heating temperature is preferably 900°C.

[0075] The total rolling reduction ratio of hot rolling is preferably 70% or more. This is because with a total rolling reduction ratio of less than 70%, the crystal grains of the hot-rolled sheet are not sufficiently made fine.

[0076] During hot rolling, finish rolling is preferably finished in the temperature region of 650 to 960°C, and the finishing temperature of hot-rolling may be in the γ region above the A_{r3} transformation point, or the α region below the A_{r3} transformation point. With the finishing temperature in hot-rolling process over 960°C, the crystal grains of the hot-rolled sheet are coarsened to deteriorate deep drawability after cold rolling and annealing. On the other hand, with a temperature of less than 650°C, deformation resistance is increased to increase the hot-rolling load, causing difficulties in rolling.

[0077] Preferably, cooling is started immediately after the end of final rolling in hot-rolling process to prevent normal grain growth and suppress AlN precipitation in the cooling step.

[0078] Although the cooling condition is not limited, the starting time of the cooling step is preferably within 1.5 seconds, more preferably 1.0 second, and most preferably 0.5 second, after the end of finish rolling. This is because when cooling is performed immediately after the end of rolling, a large amount of ferrite nuclei is produced due to an increase in the degree of over cooling with accumulated strain to promote ferrite transformation and suppress the diffusion of dissolved N in the γ phase into the ferrite grains, thereby increasing the amount of dissolved N present in the ferrite grain boundaries.

[0079] The cooling rate is preferably 10°C/s or more in order to ensure dissolved N. Particularly, when the finishing temperature of hot-rolling is the A_{r3} transformation point or more, the cooling rate is preferably 50°C/s or more in order to ensure dissolved N.

[0080] Then, the hot-rolled sheet is coiled. In order to coarsen a carbide, the coiling temperature is advantageously as high as possible. However, with a coiling temperature of over 800°C, the scale formed on the surface of the hot-rolled sheet is thickened to increase the load of the work of removing the scale, and progress the formation of a nitride, causing a change in the amount of dissolved N in the coil length direction. On the other hand, with a coiling temperature of less than 400°C, the coiling work becomes difficult. Therefore, the coiling temperature of the hot-rolled sheet must be in the

range of 400 to 800°C.

[0081] Then, the hot-rolled sheet is cold-rolled, but the rolling reduction ratio of cold rolling must be 60 to 95%. This is because with a rolling reduction ratio of cold rolling of less than 60%, a high r value cannot be expected, while with a rolling reduction ratio of over 95%, the r value is decreased.

[0082] The cold-rolled sheet subjected to cold rolling is then recrystallized and annealed. Although the annealing method may be either continuous annealing or batch annealing, continuous annealing is advantageous. The continuous annealing may be performed either in a normal continuous annealing line or in a continuous hot-dip galvanization line.

[0083] The preferable annealing conditions include 650°C or more for 5 seconds or more. This is because with an annealing temperature of less than 650°C, and an annealing condition of less than 5 seconds, recrystallization is not completed to decrease deep drawability. In order to improve deep drawability, annealing is preferably performed in the ferrite single phase region at 800°C or more for 5 seconds or more.

[0084] Annealing in the high-temperature $\alpha+\gamma$ two-phase region partially produces $\alpha \rightarrow \gamma$ transformation to improve the r value due to the development of the {111} aggregation structure. However, when $\alpha \rightarrow \gamma$ transformation completely proceeds, the aggregation structure is made random to decrease the r value, thereby deteriorating deep drawability.

[0085] The upper limit of the annealing temperature is preferably 900°C. This is because with an annealing temperature of over 900°C, redissolution of a carbide proceeds to excessively increase the amount of dissolved C, thereby deteriorating the natural aging property. When $\alpha \rightarrow \gamma$ transformation occurs, the aggregation structure is made random to decrease the r value, deteriorating deep drawability.

[0086] Furthermore, in the heating-up step in recrystallization annealing, slow heating is performed in the temperature region from 500°C to the recrystallization temperature to sufficiently precipitate AlN, and the like, thereby effectively decreasing the crystal grain diameter of the steel sheet.

[0087] The temperature region in which controlled heating must be performed is 500°C, at which AlN or the like starts to precipitate, to the recrystallization temperature.

[0088] The heating rate is preferably in the range of 1 to 20°C/s because with a heating rate of over 20°C/s, the sufficient amount of precipitates cannot be obtained, while with a heating rate of less than 1°C/s, precipitates are coarsened to weaken the effect of suppressing grain growth.

[0089] After the recrystallization annealing, temper rolling of 10% or less may be performed for correcting the shape and controlling surface roughness.

[0090] The cooling rate after soaking in recrystallization annealing is preferably 10 to 50°C/s. This is because with a cooling rate of 10°C/s or less, grains are grown during cooling to coarsen the crystal grains, thereby deteriorating the strain aging property and natural aging property. While with a cooling rate of 50°C/s or more, dissolved N does not sufficiently diffuse into the grain boundaries, deteriorating the natural aging property. The cooling rate is preferably 10 to 30°C/s.

[0091] Therefore, a cold-rolled steel sheet can be obtained, which has excellent deep drawability and excellent strain age hardenability, that tensile strength increased by press forming and heat treatment.

[0092] In the present invention, the above composition preferably further contains at least one of the following groups a to c:

Group a: at least one of Cu, Ni, Cr and Mo in a total of 1.0% or less;

Group b: one or both of Ti and V in a total of 0.1% or less; and

Group c: one or both of Ca and REM in a total of 0.0010 to 0.010%.

[0093] Element of group a: Cu, Ni, Cr and Mo are all contribute to an increase in strength of the steel sheet, and can be contained singly or in a combination according to demand. The effect is recognized by containing 0.01% or more each of Cu, Ni, Cr and Mo. However, with an excessively high content, deformation resistance at elevated temperatures in hot rolling is increased, or chemical conversion properties and surface treatment properties in a wide sense deteriorate, and a welded portion is hardened to deteriorate weld moldability. Therefore, Cu, Ni, Cr, and Mo are preferably contained singly at 1.0% or less, 1.0% or less, 0.5% or less, and 0.2% or less, respectively, and preferably contained in a combination at a total of 1.0% or less.

[0094] Element of group b: Both Ti and V are elements contributing refinement and homogenization of crystal grains, and may be added singly or in a combination according to demand. The effect can be recognized by containing 0.005% or more each of Ti and V. However, with an excessively high content, deformation resistance at elevated temperatures in hot rolling is increased, or chemical conversion properties and surface treatment properties in a wide sense deteriorate. Furthermore, there is the adverse effect of decreasing the amount of dissolved N. Therefore, Ti and V are preferably contained singly at 1.0% or less and 1.0% or less, respectively, and preferably contained in a combination at a total of 0.1% or less.

[0095] Elements of group c: Both Ca and REM are elements useful for controlling the form of inclusions. Particularly, when the stretch flanging property is required, these elements are preferably added singly or in a combination. When

the total of the elements of group d is less than 0.0010%, the effect of controlling the form of inclusions is insufficient, while when the total exceeds 0.010%, surface defects significantly occur. Therefore, the total of the elements of group d is preferably limited to the range of 0.0010 to 0.010%. This permits improvement in the stretch flanging property without causing surface defects.

[0096] The structure of the steel sheet of the present invention is described below.

[0097] Average crystal grain diameter of ferrite phase: 20 μm or less

[0098] In the present invention, the value used as the average crystal grain diameter is a higher one of the value calculated from a photograph of a sectional structure by a quadrature method defined by ASTM, and the nominal value determined by an intercept method defined by ASTM (refer to, for example, Umemoto et al.: Heat Treatment, 24 (1984), p334).

[0099] The cold-rolled steel sheet of the present invention maintains a predetermined amount of dissolved N in the product step. However, as a result of experiment and research conducted by the inventors, it was found that variations in strain age hardenability occur in steel sheets containing the same amount of dissolved N, and one of the main causes of the variations is a crystal grain diameter. In the structure of the present invention, in order to stably obtain a high BH amount and ΔTS , the average crystal grain diameter is at least 20 μm or less, and preferably 15 μm or less. Although the detailed mechanism is not known, this is supposed to be related to the segregation and precipitation of alloy elements in the crystal grain boundaries, and the influences of processing and heat history on the segregation and precipitation.

[0100] Therefore, in order to achieve stability of strain age hardenability, the average crystal grain diameter of the ferrite phase is 20 μm or less, and preferably 15 μm or less.

[0101] As described above, in the present invention, "excellent strain age hardenability" means that in aging under conditions of holding at a temperature of 170°C for 20 min. after pre-deformation with a tensile strain of 5%, the increment in deformation stress (represented by the amount of BH = yield stress after aging - pre-deformation stress before aging) after aging is 80 MPa or more, and the increment in tensile strength (represented by ΔTS = tensile strength after aging - tensile strength without strain aging) after strain aging (pre-deformation + aging) is 40 MPa or more.

[0102] In defining the strain age hardenability, the amount of pre-strain (pre-deformation) is an important factor. As a result of research of the influence of the amount of pre-strain on strain age hardenability, the inventors found that (1) the deformation stress in the above-described deformation system can be referred to as an amount of approximately uniaxial strain (tensile strain) except the case of excessive deep drawing, (2) the amount of uniaxial strain of an actual part exceeds 5%, and (3) the strength of a part sufficiently corresponds to the strength (YS and TS) obtained after strain aging with a pre-strain of 5%. In the present invention, based on these findings, the pre-deformation of strain aging is defined to a tensile strain of 5%.

[0103] Conventional coating and baking conditions include 170°C and 20 min as standards. When a strain of 5% is applied to the steel sheet of the present invention, which contains a large amount of dissolved N, hardening can be achieved even by aging at low temperature. In other words, the range of aging conditions can be widened. In order to attain a sufficient amount of hardening, generally, retention at a higher temperature for a longer time is advantageous as long as softening does not occurs by over aging.

[0104] Specifically, in the steel sheet of the present invention, the lower limit of the heating temperature at which hardening significantly takes place after pre-deformation is about 100°C. On the other hand, with the heating temperature of over 300°C, hardening peaks, thereby causing the tendency to soften and significantly causing thermal strain and temper color. With the retention time of about 30 seconds or more, hardening can be sufficiently achieved at a heating temperature of about 200°C. In order to obtain more stable hardening, the retention time is preferably 60 seconds or more. However, retention for over 20 minutes is practically disadvantageous because further hardening cannot be expected, and the production efficiency significantly deteriorates.

[0105] Therefore, in the present invention, the conventional coating and baking conditions, i.e., the heating temperature of 170°C and the retention time of 20 minutes, are set as the aging conditions. With the steel sheet of the present invention, hardening can be stably achieved even under the aging conditions of a low heating temperature and a short retention time, which fail to achieve sufficient hardening in a conventional bake-hardening steel sheet. The heating method is not limited, and atmospheric heating with a furnace, which is generally used for coating and baking, and other methods such as induction heating, heating with a nonoxidation flame, a laser, plasma, or the like, etc. can be preferably used.

[0106] The strength of an automobile part must be sufficient to resist an external complicated stress load, and thus not only strength in a low strain region but also strength in a high strain region are important for a raw material steel sheet. In consideration of this point, in the steel sheet of the present invention used as a raw material for automobile parts, BH is 80 MPa or more, and ΔTS is 40 MPa or more. More preferably, BH is 100 MPa or more, and ΔTS is 50 MPa or more. In order to further increase BH and TS, the heating temperature in aging may be set to a higher temperature, and/or the retention time may be set to a longer time.

[0107] The steel sheet of the present invention has the advantage that when the steel sheet is allowed to stand at room temperature for about one week without heating after forming, an increase in strength of about 40% of that at the

time of complete aging can be expected.

[0108] The steel sheet of the present invention also has the advantage that even when it is allowed in an unmolded state at room temperature for a long time, aging deterioration (an increase in YS and a decrease in El (elongation)) does not occur, unlike a conventional aging steel sheet. In order to prevent the occurrence of a trouble in actual press forming, it is necessary that in aging at room temperature for 3 months before press forming, an increase in YS is 30 MPa or less, a decrease in elongation is 2% or less, and a recovery of yield point elongation is 0.2% or less.

[0109] In the present invention, the surface of the cold-rolled steel sheet may be coated by hot-dip galvanization or alloying hot-dip galvanization without any problem, and TS, BH and ΔTS are equivalent to those before plating. As the plating method, electro-galvanization, hot-dip galvanization, alloying hot-dip galvanization, electro-tinning, electric chromium plating, electro-nickeling, and the like may be preferably used.

Claims

1. A cold-rolled steel sheet having excellent strain age hardenability comprising a composition, by mass %:

C: 0.15% or less;
Si: 0.005% to 1.0% ;
Mn: 0.01% to 2.0% ;
P: 0.1% or less;
S: 0.01% or less;
Nb: 0.005% to 0.050% ;
B: 0.0001% to 0.0030% ;
Al: 0.005 to 0.030%; and
N: 0.0050 to 0.0400%;

wherein N/Al is 0.30 or more, the amount of dissolved N is 0.0010% or more, optionally further comprising at least one of the following Groups a to c:

Group a: at least one of Cu, Ni, Cr and Mo in a total of 1.0% or less;
Group b: at least one of Ti and V in a total of 0.1% or less; and
Group c: one or both of Ca and REM in a total of 0.0010 to 0.010%,

and the balance is composed of Fe and inevitable impurities, and said steel sheet has a crystal grain diameter of 20 μm or less.

Patentansprüche

1. Ein kaltgewalztes Stahlblech mit exzellenter Reckalterungseigenschaft, umfassend eine Zusammensetzung, in Gew.-%:

C: 0,15 % oder weniger;
Si: 0,005 % bis 1,0 %;
Mn: 0,01 % bis 2,0 %;
P: 0,1 % oder weniger;
S: 0,01 % oder weniger;
Nb: 0,005 % bis 0,050 %;
B: 0,0001 % bis 0,0030 %;
Al: 0,005 bis 0,030 %; und
N: 0,0050 bis 0,0400 %;

wobei N/Al 0,30 oder mehr ist, die Menge an gelöstem N 0,0010 % oder mehr ist, optional ferner umfassend wenigstens eine der folgenden Gruppen a bis c:

Gruppe a: wenigstens eine von Cu, Ni, Cr und Mo in einer Gesamtmenge von 1,0 % oder weniger;
Gruppe b: wenigstens eine von Ti und V in einer Gesamtmenge von 0,1 % oder weniger; und
Gruppe c: eine oder beide von Ca und REM in einer Gesamtmenge von 0,0010 bis 0,010 %;

und der Rest besteht aus Fe und unvermeidbaren Verunreinigungen, und das Stahlblech hat einen Kristallkorn-
durchmesser von 20 μm oder weniger.

5 Revendications

1. Tôle d'acier laminée à froid ayant une excellente aptitude au durcissement par vieillissement par contrainte comprenant une composition, en % en masse :

10 C : 0,15 % ou moins ;
Si : 0,005 % à 1,0 % ;
Mn : 0,01 % à 2,0 % ;
P : 0,1 % ou moins ;
S : 0,01 % ou moins ;
15 Nb : 0,005 % à 0,050 %
B : 0,0001 % à 0,0030 %
Al : 0,005 à 0,030 % ; et
N : 0,0050 % à 0,0400 % ;

20 dans laquelle N/Al est 0,30 ou plus, la quantité de N dissous est de 0,0010 % ou plus, et elle comprend éventuellement en outre au moins l'un des groupes a à c suivants :

groupe a : au moins l'un de Cu, Ni, Cr et Mo pour un total de 1,0 % ou moins ;
groupe b : au moins l'un de Ti et V pour un total de 0,1 % ou moins ; et
25 groupe c : un ou les deux de Ca et REM pour un total de 0,0010 % à 0,010 %,

et le reste étant composé de Fe et des impuretés inévitables, et ladite tôle d'acier a un diamètre de grains cristallins de 20 μm ou moins.

30

35

40

45

50

55

FIG. 1

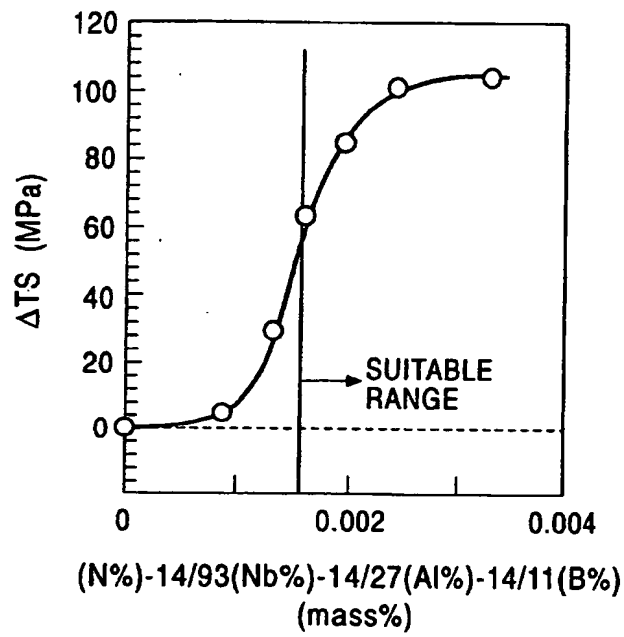


FIG. 2

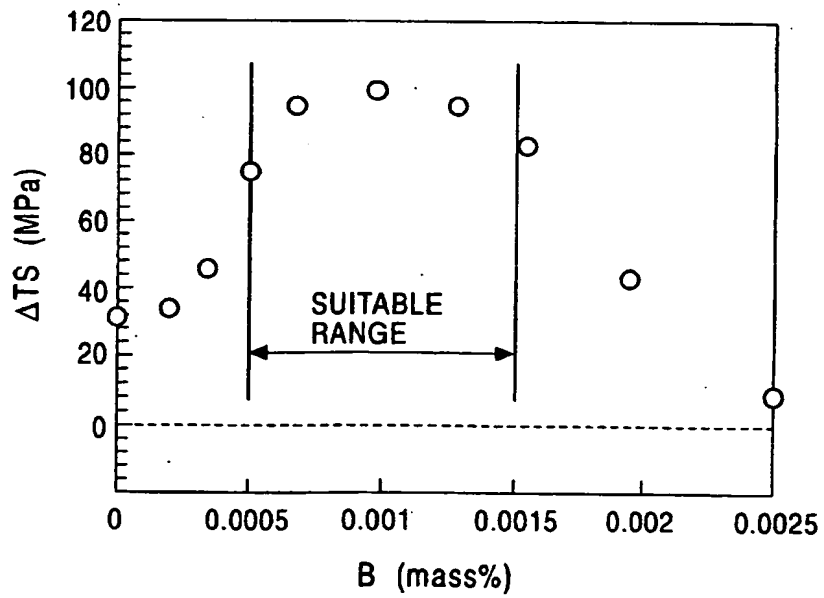


FIG. 3

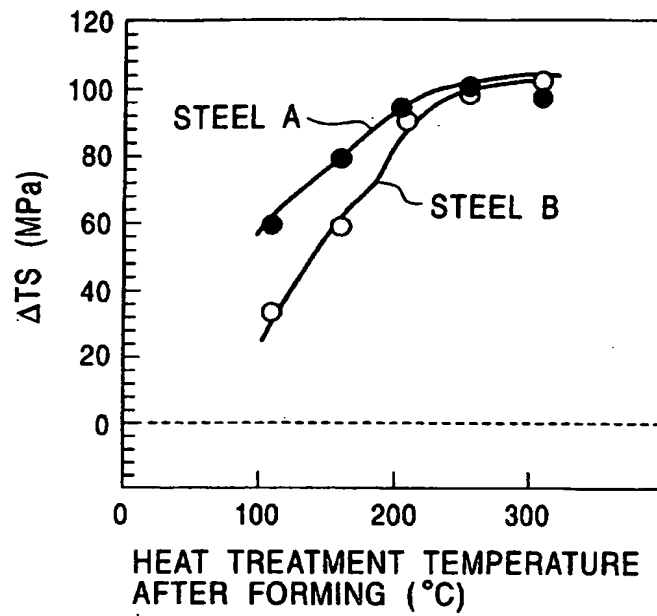


FIG. 4

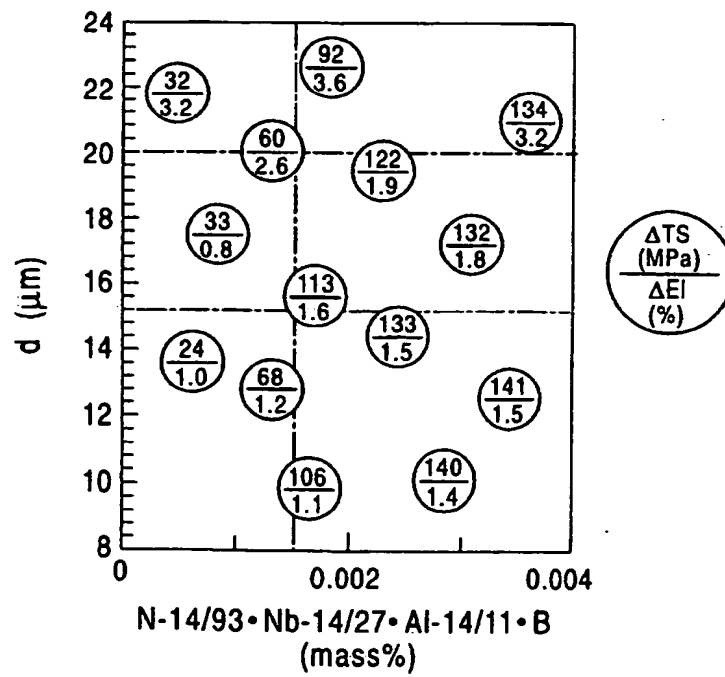


FIG. 5

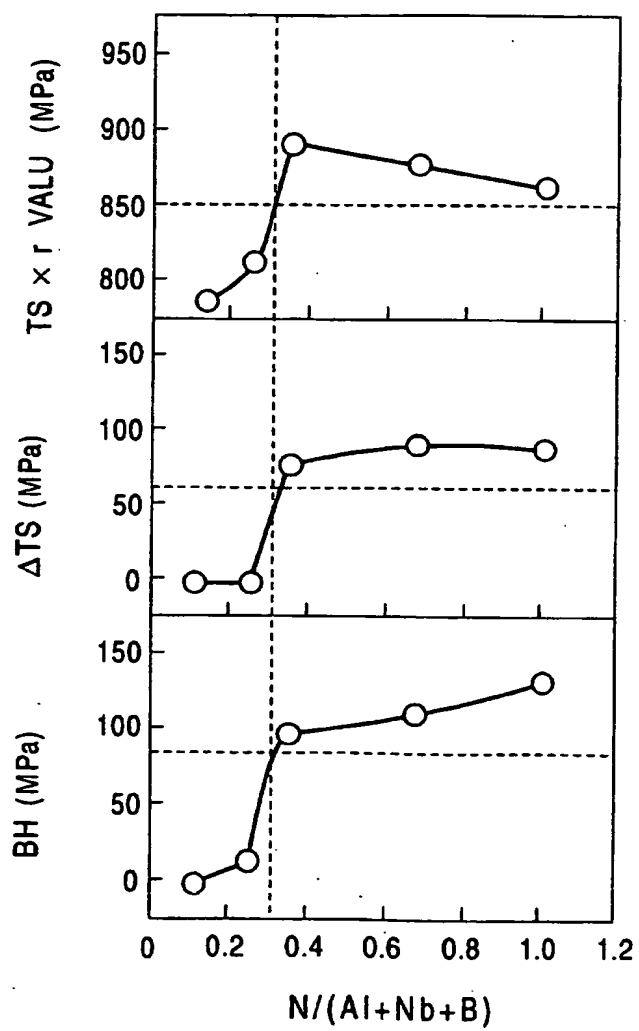


FIG. 6

